New multiphase steels with excellent machinability

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Summary

The close collaboration between Swiss Steel as steel manufacturer and Steeltec as bright bar producer led to a new family of multiphase steel products combining a tensile strength from 1000 to 1400 MPa with a minimum elongation at break of 6 to 12%, respectively.

The development of these new steels required thermodynamic modelling and laboratory scale experiments to design adequate steel concepts for conventional hot rolling and subsequent continuous air cooling. Steels with predominantly bainitic structure showed the best combination of mechanical properties and excellent machinability. These bainitic steel compositions are now regularly hot rolled and further processed to bright bars using adapted production processes.

The HSX®-family supports the trend within the automotive industry towards weight reduction and more economical processes. Conventional quenched and tempered steels can now be replaced because of the unique combination of high strength and good machinability of HSX®-steels in as delivered conditions.

Keywords: Economising part production, High strength steel, Machinability, Bainite, Thermodynamic modelling, Steel development

Introduction

Low carbon C-Mn-Si-Cr steel with a carbide free bainite-martensite microstructure can be obtained under air-cooling conditions. Such steels show good combinations of strength and toughness. To get the right properties from hot rolling the microstructural transformation kinetics must be adapted to the obtainable cooling rates of the mill. It has to be taken into account that cooling rates will vary from the centre to the surface of a bar. The chosen steel composition should not only work for one bar dimension but for a range of dimensions. A certain “robustness” against cooling variations in the range of 0.1 to 2.0 K/s is required and in this sense the alloy composition must be properly designed.

For the machining industry also good machining performance is an important part of the customers requirements. In the present development it was decided to allow the addition of sulphur to the steel as machining agent. Typical aluminium killed or boron microalloyed steel concepts therefore couldn’t be taken into consideration.

Alloy design

Assuming fully bainitic steel structures a multi-linear regression formula as reported by Mesplont et al [1] was applied to estimate the tensile strength and to adjust the chemical composition of promising candidates. Steel A – F as given in table 1 are supposed to be dominantly bainitic. The as calculated values of tensile strength after hot rolling are within a narrow window just above 1’000 MPa.

	C	Si	Mn	Cr	Mo	Rm
Steel A	0.15 %	1.2 %	1.5 %	1.2 %	0.3 %	1017 MPa
Steel B	0.18 %	1.2 %	1.5 %	1.2 %	0.2 %	1009 MPa
Steel C	0.19 %	1.2 %	1.4 %	1.2 %	0.3 %	1041 MPa
Steel D	0.19 %	1.0 %	1.4 %	1.5 %	0.3 %	1042 MPa
Steel E	0.24 %	1.0 %	1.4 %	1.5 %	0.15 %	1034 MPa
Steel F	0.25 %	0.8 %	1.2 %	1.8 %	0.15 %	1027 MPa

Table 1 steel compositions in wt.-% and corresponding tensile strength values

In order to find out the conditions where ferrite formation should be most successfully suppressed variations in steel composition were investigated on a computer applying thermodynamic phase transformation theory. The applied model as developed by Bhadeshia [2] calculates the time required to initiate the allotriomorphic ferrite (upper C-curve) and of Widmannstätten ferrite and bainite (lower C-curve). The curves represent the beginning of isothermal transformation at any temperature. The formation of Widmanstätten ferrite requires an undercooling below the equilibrium temperature such that the magnitude of the driving force is greater than 50J/mol in order to allow for the strain energy due to displacive transformation. Using the bainite und martensite start temperatures as upper and lower limits for bainite formation the model calculates the TTT diagram from the chemical composition of the steel (in the elements C, Si, Mn, Cr, Mo, Ni, V).

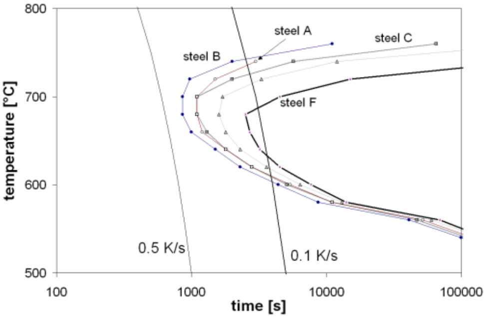


Fig. 1 zero fraction lines of allotriomorphic ferrite

In fig. 1 the calculated allotriomorphic ferrite zero fraction lines are plotted for five steel compositions (steel E is not shown because steel E and F are almost falling together). The trajectories of continuous cooling (starting from 1’000 °C) at 0.1 and 0.5 K/s are given.

Although these results are strongly idealized (for example the influence of hot deformation and also the influence of cooling rate on the phase transformation are not taken into account) they show the tendencies. At a cooling rate of 0.1 K/s ferrite formation is best suppressed in case of steel F.

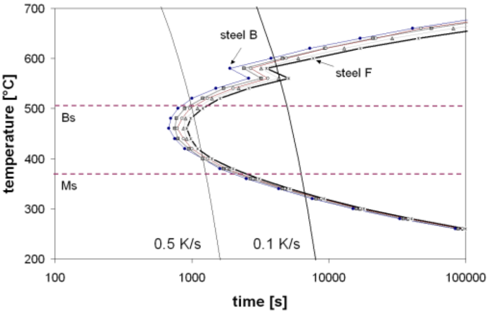
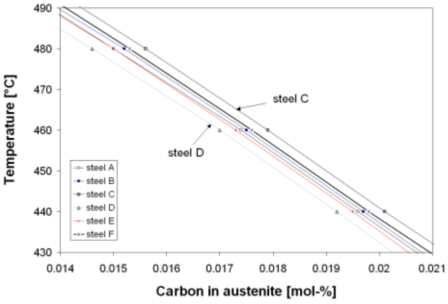


Fig 2 shows the zero fraction lines of the “lattice distorted ferrite” (here Widmanstätten ferrite). To obtain a better estimate of the bainite region the calculated bainite and martensite start temperatures [3] of steel F are given as dashed lines.

Fig. 2 zero fraction lines of Widmanstätten ferrite

There seems to be little difference between the investigated six steel compositions. It must be considered that the transformation from austenite to bainite cannot be complete. During bainite formation carbon moves into the untransformed austenite. At a certain carbon concentration austenite and bainite will have the same free energy and the bainite formation stops.

This “incomplete reaction phenomenon” is described thermodynamically by the T_0’ concept [4-8]. The T_0’ curve is the locus of points, on a temperature versus carbon plot, where austenite and bainite have the same free energy. The stored energy of the bainite due to the lattice distortion is taken as 400 J/mol.



The T_0’ concept can be used to estimate which steel would form more bainite if the carbon contents of the steels are close together. The comparison between steels B, C and D indicates that steel C tends to form more bainite than B and D.

Fig. 3 *T_0’ curves of the designed steels*

From thermodynamic considerations the tendency to form allotriomorphic ferrite seems to be the most important difference between the investigated steel compositions. Steel E and F are most promising candidates to suppress ferrite at low cooling rates. If intermediate cooling can be used than steel A (with low carbon content) or steel C (with upper T_0’ curve) could be attractive alternatives.

Laboratory experiments

Before going into large scale industrial production 25 kg ingots of $80x80mm^2$ were cast and hot rolled at the “Institut für Metallumformung” in Freiberg.

The production of bainitic steels at low cooling rates is challenging in the sense of avoidance of undesirable ferrite formation. Investigations were therefore done with steel compositions D-F at a cooling rate of 0.3K/s. These steels were hot rolled at 1200°C to 23 mm round bars. The bars were put into a temperature controlled furnace after rolling to simulate the right cooling conditions. The cooling curve was measured with thermocouples on a reference probe (fig. 4).

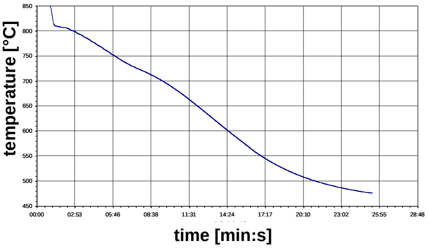


Fig. 4 Measured cooling curve between 800°C and 500°C. The cooling rate was 0.27 K/s.

Metallographic analysis of the microstructure of the three steels was done after LePera etching [9]. The pictures as shown in fig. 5 were taken on longitudinal samples (position R/2). Martensite appears with a brown colour, ferrite is white and the rest (blue) is mainly bainite (neglecting the very small islands of retained austenite which are not visible here).

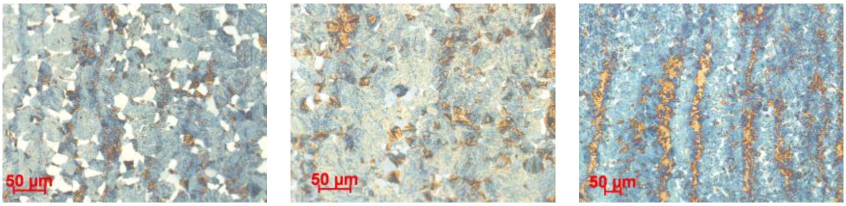


Fig 5 microstructure of steel D (left), E(middle) and F(right)

Steel D exhibits a non-negligible amount of ferrite grains. This is in good agreement with the thermodynamic calculations. The surroundings of the ferrite are expected to be enriched by manganese. For this reason the ferrite is partly decorated by martensite.

The transformation of austenite to martensite starts at ≈ 400°C and is not totally completed at room temperature. During transformation (from 400°C to ambient temperature) elements like carbon and hydrogen will be captured in unfavourable positions within the steel lattice. Kinetics will be slow due to the low temperature and the steel is expected to age over a long period of time. In this work all specimen in the as hot rolled state were therefore tempered in an inert atmosphere for one hour at 300°C to accelerate the ageing before testing. The results of the tensile tests after tempering were shown in table 2. The tensile strength level is about 100 MPa higher as calculated for the fully bainitic structure.

Considering the presence of a small amount of martensite the agreement between calculation and measurement is fairly good. The higher tensile strength of steel F indicates the presence of more martensite. E is favoured due to its microstructure.

	Rp0.2	Rm	A5	Z
	MPa	MPa	%	%
Steel D	**804**	**1100**	**13.3**	**42.5**
Steel E	**843**	**1158**	**11.9**	**35.4**
Steel F	**915**	**1207**	**11.5**	**42.1**

Table 2 results from tensile tests

Industrial trials

From the above investigations steels A (severe cooling, i.e. small diameters) and E (slow cooling) were the favourable steel compositions. They were tested in industrial trials under regular production conditions. Three heats of steel A and two heats of steel E were produced in an electric arc furnace from scrap and cast on a continuous casting machine to 140x140 mm^2 or 150x150 mm^2 billets. In the hot rolling mill the billets were reheated to 1200°C and conventionally hot rolled to wire rods (13 and 17.5 mm) or bars (22 and 53 mm).

To study the influence of cooling parameters on the microstructure and on the mechanical properties wire rods were manufactured from steel A on the Stelmor conveyor line. This production route allows to apply high cooling rates and to reach fine microstructures.

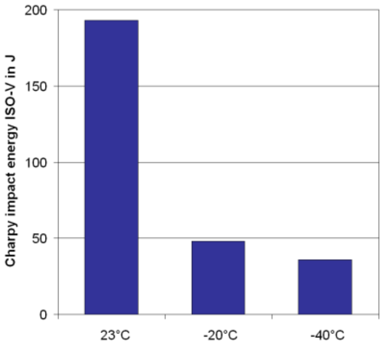


Fig. 6 Charpy impact energies of a 17.5 mm wire rod. Cooling rate ~3 K/s.

Excellent combinations of tensile and impact strength were obtained at cooling rates above 2K/s. The ISO-V impact energies determined on a 17.5mm wire rod are given in fig. 6. In this case the mechanical properties from tensile testing were:

Rp0.2 = 1001 MPa
Rm = 1135 MPa
A5 = 15 %
Z = 68 %

Sufficient high cooling rates are essential to get the fine microstructure which is the precondition for good toughness. Decreasing cooling rates impact energy values are falling. In bar production (cooling rates between 0.3 and 1.0 K/s) the impact energies were always below 100 J.

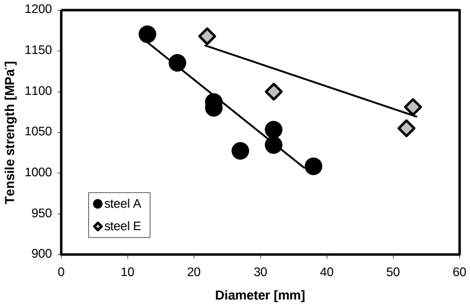


Fig. 7 Tensile strength as a function of diameter. Diameter below 20 mm are wires. The lines are only guides to the eye. No linear relationship is expected.

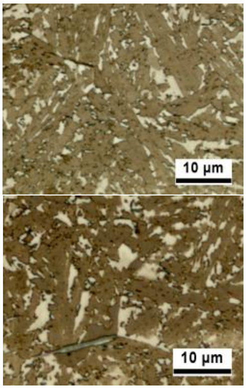


Fig. 8 Micrographs of a 32 mm bar: steel A (above) and steel E(below)

Bars from 20 mm to 53 mm in diameter were produced with both steel compositions. The tensile strength values (including the two wire rod trials) are shown in fig 7. The tensile strength falls with increasing diameter (i.e. decreasing cooling rate). This drop is more pronounced for steel A. For this steel the target value of >1000 MPa cannot be held safely if the diameter exceeds 30 mm. Steel E fulfils this criterion over the whole diameter range. The microstructures of both steels are compared in fig. 8 for a 32 mm bar.

Using the hot rolled samples Steeltec AG manufactured bright bars. As an example the mechanical properties of steel E in the drawn, straightened and stress relieved state are given in table 3. The 0.2-% offset yield stress increases dramatically during drawing.

	Rp0.2	Rm	A5
20 mm	1362 MPa	1372 MPa	8.7 %
50 mm	1313 MPa	1338 MPa	6.9 %

Table 3 mechanical properties of bright bars from steel E

Applications

Now different bainitic steels of the above described C-Mn-Cr-Si type are in use. Today’s standard steel contains 0.15 % sulphur and has a chemical composition which is laying between steel A and E. The non-conventional bright bar production process is part of the success of this new product family with the name HSX$^{®}$ (**H**igh **S**trength e**X**cellence). The most commonly used products are HSX$^{®}$130 and HSX$^{®}$Z12. Since two years both products are commercially available within a diameter range from 18 to 50 mm. Their mechanical properties are given in fig. 9.

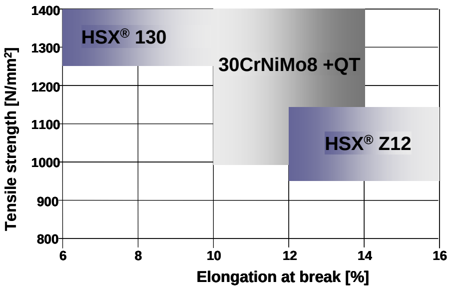


Fig. 9 Mechanical properties of HSX®130 and HSX®Z12 in comparison to 30CrNiMo8 +QT

HSX®130 is applied in areas where high forces have to be transferred (for example in gear shafts, camshafts, hydraulic cylinders or pneumatic cylinders).

HSX®Z12 is preferentially used if the high load transmission is accompanied by dynamic stresses.

Due to its high strength of min. 1250 MPa the as drawn material HSX®130 is an appropriate candidate for use in applications transferring high forces. For parts with alternating loads HSX®Z12 could be an alternative due to its higher elongation value of min. 12%. Both, HSX®130 and HSX®Z12, replaced conventional QT steel grades in some applications like electric adjustment mechanism, hydraulic components, components in safety systems and for camshafts.

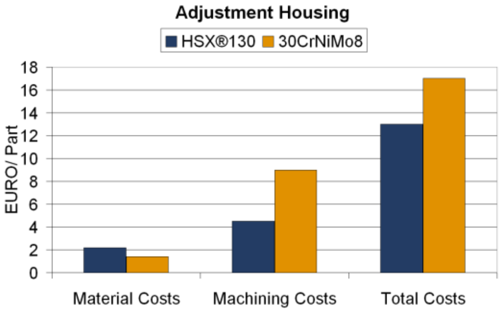


Fig. 10 shows a cost estimation of a adjustment housing (clutch) out of HSX®130 in comparison to the same component out of 30CrNiMo8. Material and machining costs (50-60% of total cost) are given in detail.

Fig. 10 Part production costs of a hydraulic component

Even with higher material costs, savings were made. A significant saving over the total process route is achievable by designing the process without any additional heat treatment.

Conclusions

Thermodynamic models and laboratory test equipment were routinely applied to accelerate the development of mass-tailored steel grades. New bainitic-martensitic high-strength steels of the HSX-family exhibiting improved machinability were presented. They have the potential to replace standard QT grades and to economise part production by eliminating heat treatment and subsequent process steps.

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References

1. C. Mesplont, T. Waterschoot, S. Vanderputte, D. Vanderschueren and B.C. De Cooman, 41th MSWP Conf. Proc., ISS, pp 515-524 (1999)

2. H.K.D.H. Bhadeshia, "www.msm.cam.ac.uk/map/steel/programs/mucg46-b.html", Materials Algorithms Project (MAP), University of Cambridge, Dept. of. Mat. Science and Metallurgy

3. W. Steven, and A.J. Haynes, JISI 183, pp. 349-359 (1956)

4. H.K.D.H. Bhadeshia, Acta Metall., 29, pp 1117- (1981)

5. H.K.D.H. Bhadeshia and A.R. Waugh, Acta Metall., 30, pp 775- (1982)

6. L.C. Chang and H.K.D.H. Bhadeshia, Mater. Sci. Eng. A, A184, L17 (1994)

7. I. Stark, G.D.W. Smith and H.K.D.H. Bhadeshia, Proc. Of Int. Conf. Phase Transformation, ed. by G.W. Lorimer, Institute of Metals, pp 211- (1988)

8. F.G. Caballero, M.J. Santofimia, C. Capdevila, C. Garcia-Mateo and C. Garcia de Andrei, ISIJ Int., Vol 46, No 10, pp 1479-1488 (2006)

9. E. Girault, P. Jacques, Ph. Harlet, K. Mols, J. Van Humbeeck, E. Aernoudt, F. Delannay, Materials Characterization 40, pp 111-118 (1998)